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The Metallurgist.

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Microstructure and Properties.

MERE inspection of the microstructure of a specimen of metal affords no information about the properties of the material. It is essential that the observer should have a background of knowledge of the constitution of the alloy system involved. Hence the study of microstructures has, for the past fifty years, been linked up to an increasing extent with those investigations of other physical properties comprised under the term metallography. It is just fifty years ago since Sorby announced, at a meeting of the Iron and Steel Institute, his discovery of the "pearly constituent" in steel. This was long before the time when the origin of eutectic and eutectoid constituents was properly understood. Sorby was interested in the morphological aspect of metallography and in the analogies between the structures of metals and of rocks. Others since have been mainly interested in the same aspect of metallic structures. Among them was H. M. Howe. It is true that in the course of his long career he showed an increasing consciousness that a knowledge of equilibrium conditions was a vital factor in the interpretation of his work, and he himself made notable contributions to the study of the constitution of alloys, but still his fame as a metallographer rests mainly on the fact that he was one of the pioneers in the correlation of the macro and microstructure of metals with their mechanical properties and physical defects.

It was therefore appropriate that the Thirteenth Howe Memorial Lecture of the American Institute of Mining and Metallurgical Engineers, delivered in New York by Professor H. F. Moore on February 20th, should have dealt with "Correlation between Metallography and Mechanical Testing." The lecturer showed how such correlation has already been achieved, or is being actively pursued, in connection with heat treatment in the light of constitution and grain size and with the mechanism of elastic failure, creep, and fracture of metals. The lecture contained a plea for increased activity in the joint study of these two branches of the science of metals, since the greater the number of links which can be established between them, the firmer will be the foundation on which future theory and practice can be built.

It is a curious fact that although the mathematical theory of a homogeneous elastic material furnishes a satisfactory basis for design, metallographists have shown that metals are in fact not homogeneous, but are made up of crystal grains of various orientation and different strength and, in addition, may contain non-metallic inclusions. Their elastic behaviour is "statistically" accurate, and their average behaviour is elastic in spite of the imperfect elasticity of the individual grains. Professor Moore dealt at some length with the importance of the slight plastic action by which localised stresses are relieved and machine parts enabled to hold together, and so led up to the paradox, "The reason why we can use the theory of elasticity for designing actual structures and machines is that our actual materials are not perfectly elastic."

In the course of the lecture Professor Moore pointed out several desirable fields of correlative study which require the use of very high magnifications. In this direction the work of Mr. Lucas represents the best achievement so far. There is no doubt that many problems could be attacked with the aid of greater resolving power in the microscope: ageing, precipitation-hardening, temper-brittleness, inter-crystalline impurities, veining, and the structure and strength of the grain boundary. The demand for increased resolution naturally suggests ultra-violet light, but attempted applications of ultra-violet light microscopy to metals have, so far, not been very successful. The resolution secured by short wave-length is in part lost by the low aperture of the usual ultra-violet light objective. A high-aperture metallurgical objective for short wave-length visible light, such as that developed by the British Scientific Instruments Research Association, may well lead more quickly to important practical advances, since it can be used with ordinary photomicrographic equipment. High power work is for the relatively few who can afford the time and the expense involved. On the other hand, the importance of macrostructure and of micrography at moderate magnification is apparent to all workers in metals, and much remains to be done even in these fields for further study of the relative seriousness of various types of imperfection of structure in relation to practical behaviour in test or in use.

The Diffusion of Gases Through Metals.

SINCE the appearance of a previous article[1] summarising the work of Dr. Smithells and Mr. Ransley, the fact that nitrogen will not diffuse through copper has been confirmed by de Boer and Fast,[2] who, at the same time, obtained evidence that water vapour would diffuse through copper at high temperatures, its rate of diffusion at 800 deg. Cent. being slower than that of hydrogen. The results of an investigation by Baukloh and Kayser[3] on the permeability of

[1] THE METALLURGIST, October 25th, 1935, page 68.

[2] J. H. de Boer and J. D. Fast, *Rec. Trav. Chim. Pays-Bas*, December 15th, 1935, 54, 970.

[3] W. Baukloh and H. Kayser, *Zeit. Metallk.*, December, 1935, 27, 281.

certain metals and alloys to hydrogen have also been published. Only a qualitative relationship was traced between solubility and permeability. Hydrogen was found to diffuse through iron, nickel, nickel-iron, and nickel-copper alloys at temperatures above 550 deg. Cent., nickel being the most permeable metal. Nickel was, however, not permeable to argon and the other inert gases. This is in general accordance with Smithells and Ransley's observations, but Baukloh and Kayser further stated that copper and aluminium tubes were impermeable to hydrogen. As regards copper, this is not in accordance with previous work. As regards aluminium, the results might have been accepted as probably correct before the publication in the "Proceedings" of the Royal Society for November 15th, 1935, of a further communication by Dr. C. J. Smithells and Mr. G. E. Ransley (both of the research staff of the General Electric Company), dealing with the diffusion of hydrogen through aluminium. This subject had been previously investigated by Deming and Hendricks,[4] who came to the conclusion that aluminium was impermeable to hydrogen at all temperatures up to 550 deg.; but Smithells and Ransley point out that very small rates of diffusion would not be detected by Deming and Hendricks' experimental method.

The solubility of hydrogen in aluminium has been measured by a number of workers,[5] who have found a distinct solubility in the molten metal, but nothing measurable in the solid state. The gas found on heating aluminium *in vacuo* often contains as much as 80 per cent. of hydrogen. Its origin is now generally supposed to be the decomposition of water vapour (in the furnace gases, &c.) by the hot metal and subsequent absorption of the gas. When the metal solidifies some of this gas is retained in fine bubbles.

In their previous paper[6] Smithells and Ransley showed that the first process in diffusion was the absorption of a film of gas on the surface of the metal and might be markedly influenced by the presence of an oxide film. According to Pilling and Bedworth a film of oxide 0·00002 cm. thick is quite impervious to oxygen, and as ordinary aluminium is always covered with such fine film the present workers considered that the failure to demonstrate diffusion or solid solubility might be due to the action of this film in preventing diffusion of the gas into, or out of, the metal. They therefore decided to determine the diffusion of hydrogen through aluminium (1) in the normally tarnished condition, (2) after anodic oxidation, and (3) with the oxide film removed by abrasion *in vacuo*.

The material used in this work was commercial aluminium bar, which was drawn into tubing having an external diameter of 3·2 mm. and a wall thickness of 0·45 mm. A piece 30 cm. long was closed at one end by welding and the other end was soldered to a copper/glass joint as previously described. Analysis showed the metal to have the following composition:

	Per cent.
Copper	0·03
Iron	0·90
Manganese	0·03
Silicon	0·40
Zinc	0·17

The apparatus used was as previously described with the exception that an arrangement was introduced for scraping both the inside and outside surfaces of the tube *in vacuo* inside the diffusion apparatus. Scratch brushes of fine steel wire bearing with great pressure on the surface were attached to blocks of iron by nickel tubes and these could be moved horizontally by a powerful solenoid, and either the outside or inside surface scratched independently. Owing to the presence of the scraper and the consequent difficulty of attaching leads to the tube, the direct method of heating the tube was abandoned and a tungsten helix between two concentric fixed silica tubes was used.

A series of six experiments made by the authors is particularly interesting. These experiments were all made at 580 deg. Cent. and at a pressure of hydrogen of 29 cm. It was found that under these conditions enough gas diffused in thirty minutes to enable an accurate estimate of the rate to be made. The rate of diffusion D is expressed as the volume of gas in cubic centimetres at N.T.P. diffusing per second through 1 square centimetre of surface of 1 mm. thickness.

First Run.—Tube run in hydrogen for three hours before observations were made. $D=2\cdot0\times10^{-7}$, but diminished slightly with time.

Second Run.—Tube allowed to cool, and the outer surface vigorously scratched. It was then again heated to 580 deg., and at the end of the first hour $D=3\cdot8\times10^{-7}$, but was definitely diminishing.

Third Run.—The outer surface of the tube was scratched again and the average rate of diffusion determined over half-hour periods. The initial rate was now $4\cdot2\times10^{-7}$, but fell in five hours to a constant value $1\cdot4\times10^{-7}$.

Fourth Run.—The outer surface was scratched again, giving an initial rate $4\cdot0\times10^{-7}$, which fell in seven hours to $0\cdot9\times10^{-7}$.

Fifth Run.—Both the inner and outer surfaces of the tube were scratched. The average rate during the first half-hour was now $6\cdot3\times10^{-7}$.

These experiments show conclusively that hydrogen diffuses through aluminium at 580 deg.; in fact, the rate is the same order of magnitude as found by the authors for copper at the same temperature. The rate may, however, vary considerably with the condition of the surface. Although the initial rate was found to vary considerably, apparently after about ten hours the surface returns to a definite state for which the rate is about $0\cdot8\times10^{-7}$. The authors consider that this condition corresponds to a definite degree of oxidation of the surface, and that, although the hydrogen was pure, some of it would be converted by contact with the hot tube into atomic hydrogen, which would reduce the oxidised portions of the nickel wire even in the cold. The water vapour formed in this way would oxidise the surface of the metal. For the sixth run the tube was removed from the apparatus, polished, and anodically oxidised in a 3 per cent. chromic acid solution. The initial rate was $1\cdot33\times10^{-7}$, and this fell in $2\frac{1}{2}$ hours to $0\cdot9\times10^{-7}$. These rates were of the same order as the final values obtained in previous runs, and support the conclusion that the decrease is due to oxidation of the surface of the metal.

TABLE I.—*Effect of Pressure on Rate of Diffusion at 558 deg. Cent.*

Pressure (mm.).	Rate D.
32·5	$0\cdot18\times10^{-7}$
95	0·32
150	0·42
290	0·59

[4] H. G. Deming and B. C. Hendricks, *Journ.*, Amer. Chem. Soc., 1923, 45, 2857.

[5] K. Iwasé, Sci. Rep. Tôhoku Imp. Univ., 1926, 15, 531; P. Röntgen and H. Braun, *Metallwirtschaft*, 1932, 11, 459; P. Röntgen and F. Möller, *Metallwirtschaft*, 1932, 11, 685 and 697; L. L. Bircumshaw, "Trans.," Faraday Soc., 1935, 35, 1439.

[6] "Proc.," Roy. Soc., 1935, 150, 172.

The effect of pressure was studied at 558 deg. for a tube in the fully decayed condition ($D=0\cdot8\times10^{-7}$ at 29 cm. pressure). The results given in Table I can be represented by the usual equation :

$$D=k\sqrt{P}.$$

The effect of temperature was determined for a scraped tube having the high rate and also for an anodically oxidised tube. The high rate ($5\cdot0\times10^{-7}$ at 580 deg. and 29 cm. pressure) obtained with the same tube used for the previous measurements was on this occasion maintained for seven hours without showing any marked decay, although the experimental conditions were normal. The results are given in Table II and the usual experimental equation :

$$D=k\,e^{-b/T}$$

gives the effect of temperature on the rate of diffusion. Consistent results were difficult to obtain with the anodically oxidised tube owing to the low rate of diffusion at any but the highest temperatures.

Table II.—*Effect of Temperature on Rate of Diffusion at a Pressure of 29 cm.*

Temp., deg. abs.	Rate D.
782	$1\cdot09\times10^{-7}$
793	$1\cdot17$
805	$1\cdot75$
818	$2\cdot33$
842	$4\cdot09$
865	$5\cdot92$
878	$9\cdot70$

Smithells and Ransley conclude, from the fact that the usual diffusion equation represents the effect of temperature and pressure on the rate of diffusion, that the gas is dissociated at the metal surface. They consider that under the best experimental conditions it is impossible to avoid contamination of the surface with oxygen, and this, together with the high temperature coefficient of diffusion produces extremely slow rates of diffusion at temperatures more than 100 deg. below the freezing point of the metal. The fact that hydrogen does diffuse through solid aluminium shows that there must be some solubility of the gas in the metal. They consider that the failure of previous investigations to detect this solubility is due to the slow rate of diffusion under the ordinary conditions of experiment.

The work of Dr. Smithells and Mr. Ransley now reported is again notable for the high degree of experimental skill which has enabled the authors to establish the difficult subject of the diffusion of hydrogen through aluminium on a sound quantitative basis, and, incidentally, to contribute a most valuable study of the conditions of formation, and effect on behaviour, of the surface film on aluminium.

Influence of Working Conditions on the Porosity of Tinplate.*

By W. E. HOARE, B.Sc.†

In the work of Peter and Le Gal, recently published under this title, the effects of the following factors were examined :—Tin yield, double tinning, time of pickling, treatment in palm oil, porosity of upper and lower sides of the sheet, flux, wetness of sheets, and surface damage after tinning.

Although attention was paid to the elimination of influences which might be expected to cause porosity, no pore-free tin-plate could be obtained. Again, on examining numerous commercial tinplates, with tin yields of from 28–52 g/m², and originating from several different countries, no samples were found to be free from pores. The authors conclude that the conditions necessary for obtaining substantially pore-free coatings with tin yields of reasonable thinness are not yet fully known.

In the experimental investigation, tin-plates having fairly normal coatings were used, the basis sheets being specially selected to give a uniform starting material.

Porosity figures were obtained by the method of Eisenkolb,[1] a ferricyanide jelly test which appears to be favoured in Germany. W. H. Creutzfeld, in the discussion, criticises the hot water test, and suggests that clearer results can be obtained with a ferricyanide-sodium chloride-agar solution absorbed on filter paper. There seems to be some risk here that clarity of record is being put before absolute accuracy, since the short times recommended may result in non-registration of some of the smaller pores, which is borne out by the fact that porosity figures quoted in the paper are in all cases very much less than would be expected. A further point in connection with porosity testing concerns the preparation of the specimen. The procedure of ten minutes' immersion in trichlorethylene followed by vigorous rubbing with leather (suggested by Creutzfeld), appears to be altogether too drastic. The tendency in Great Britain and the United States is to use either the modified ferricyanide paper test or the hot water test, preceded by cathodic degreasing or by extraction above a boiling solvent.[2] These methods have been found convenient in commercial use, and are not open to any serious theoretical objection.

Effect of Thickness of Tin Coating.

The authors find that porosity decreases with increase of tin yield. Experience in Great Britain, extending over a very large number of specimens, is in agreement with this contention. It is considered by manufacturers and consumers here that porosity values improve fairly rapidly as the tin yield is increased to 1 lb. 12 oz. per basis box (19 g/m² approximately). Higher tin yields result in a steady, but less rapid, proportionate improvement.

Double Tinning.

Peter and Le Gal find that retinning only decreases porosity by virtue of any improved tin yield that may be arbitrarily obtained by difference in setting of the tinning rolls. This is in accord with experience, where it may be said generally that the steel sheet, on entering the tin pot, has a property of producing a certain normal porosity with a fixed tin yield. The porosity, for example, of retinned menders is not generally better than that of the ordinary single-tinned sheet.

Effect of Occluded Hydrogen.

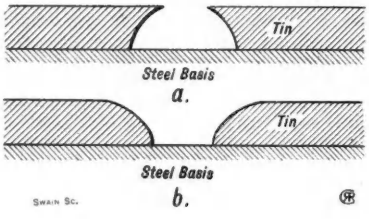
The authors have examined the possibility of the effect of evolution of hydrogen, picked up in pickling, on porosity. This conception has had much support on the Continent in the past, although there has been

* Review abstract and discussion of a paper by Fritz Peter and Georges Le Gal in *Archiv für das Eisenhüttenwesen*, December, 1935, 1935/36, 9, 285.

† Research Investigator, the International Tin Research and Development Council.

no real basis for it. The present experiments indicate that the effect, if present at all, is small, and of rather less importance than some other factors. Increase of pickling time, which, if the effect is present would be expected to increase discontinuity, caused a marked improvement in the coating. The increased tin yields obtained with the longer pickling times were not in themselves sufficient to account for the improved porosity. The improvement was rightly attributed by the authors to increased cleanness of the surface with respect to scale and non-metallic inclusions. In discussion, the only dissentient agreed that positive evidence for the effect of hydrogen evolution on porosity was lacking, but criticised the results on the grounds that an inhibitor had been used which was specifically designed to reduce hydrogen absorption.

The statement that the shape of pores indicates that they are due to gas bubbles is not confirmed by the work of Hoare and Chalmers.[2] It would be expected that an issuing gas bubble would cause a pore shape roughly as shown in Sketch (*a*), while



it has been shown[3] that actually normal pores are of the shape indicated in (*b*).

Interesting results were obtained when zinc chloride solution was used as a pickling agent. A clean, metallic surface was obtained, presumably without the possibility of hydrogen occlusion, by treating for three to fifteen minutes in zinc chloride solution at 130 deg. Cent., but this led to no improvement in porosity. Other experiments, however, indicated that hydrogen occlusion may have some effect. When opportunity for the escape of hydrogen was given, for example, by storage before tinning, or by immersion, after pickling, in heated zinc chloride solution, a decrease of porosity was obtained.

EFFECT OF IMMERSION IN PALM OIL.

In search of further information on this point, a series of experiments was carried out in which tin-plates were immersed in palm oil at 250 deg. Cent. for varying times, and then retinned. A considerable improvement resulted and was attributed to the fluxing and washing action of the oil. It was further found that unpickled, box-annealed sheets could be satisfactorily tinned after immersion for ten minutes in palm oil at 250 deg. Cent. Palm oil, of course is used as a flux in Great Britain in the manufacture of high-grade charcoal tin-plates.

INFLUENCE OF SURFACE CONDITION.

It was established that pore sites coincided with the presence of steel surface defects recognisable on the untinned sheets. Although the formation of $FeSn_2$ would necessarily be inhibited at such points, the surface defects illustrated were apparently of such size that the sheets would in any case have been eliminated before the tinning operation.

INFLUENCE OF FLUX.

Results obtained by Peter and Le Gal showed that the porosity of the upper and lower surfaces of the sheets were not widely variant. Such a conclusion is not borne out by experience in Great Britain, and in this connection, W. Schneider, in discussion, mentioned gross differences between the porosity of the upper and lower sides. This difference has been attributed by the reviewer to the difficulty of escape of flux reaction products from the lower side as compared with the upper side.

It was further established that, provided the flux was reasonably free from contamination, the time of usage did not affect porosity. This cannot be strictly true, since the reaction

$$Zn\,Cl_2 + 2\,H_2O \longrightarrow Zn\,(OH)_2 + 2\,HCl$$

must account, sooner or later, for a decreased activity. The amount of water on the sheets entering the flux pot was found, however, to affect porosity to a marked degree, as would be expected from the above reaction. Three series of tests were carried out. Series (*a*), in which the sheets were stood vertically for fifty seconds after taking out of the water-bosh; series (*b*), in which the sheets were stood for ten seconds, and series (*c*), in which the sheets were transferred directly from the water-bosh to the tin pot. The porosity decreased to a marked degree from (*a*) to (*c*), that of (*c*) being of the order of one-tenth of that of (*a*).

The bad effect of cleaning, rough handling, and fabrication, on porosity are mentioned at the end of the paper.

Peter and Le Gal summarised their work in the following words (freely translated):—

"It was established that the porosity of tin-plate could not be attributed to any single cause. It depends principally upon the surface cleanliness of the untinned sheet and on the thickness of the coating; and further, upon the efficiency of the pickling process, the flux, &c. Hydrogen, taken up by the sheet during its manufacture, and especially during pickling, and partially set free again in the tinning bath, does not, according to the experimental results, play the dominating *rôle* in the origin of pores which is frequently attributed to it. It was further established that with the tin deposits usual to-day on commercial tin-plates, in spite of the use of all measures for the diminution of porosity, freedom from pores cannot be achieved. Pore-free tin-plate is only obtained with tin yields above 100 g/m², and only then if all available means for decreasing porosity are used."

The International Tin Research and Development Council is actively engaged in researches dealing with the porosity of tin coatings, and a paper will be published shortly dealing with the origin of porosity. Further researches are connected with detailed discussion of many of the points raised by Peter and Le Gal, and other investigators.

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Hydrogen as the Cause of Low Elongation and Reduction of Area of Steel.

THERE have been conflicting opinions for many years as to the advantages of very prolonged tempering treatments on the elongation and reduction of area of steels. On the one hand, it has been held that no advantage is gained by a prolonged tempering treatment which cannot be obtained by a shorter treatment at a higher temperature, and that any gain in ductility is offset by a corresponding lowering in yield point and maximum load. On the other hand, the advantages of very prolonged treatment at quite low temperatures have been advocated, and even rest at atmospheric temperature after rolling has been shown to improve ductility in the tensile test. The investigations of F. Körber and J. Mehovar* showed that the mechanical properties of newly rolled open-hearth steel rails varied considerably, according as to whether the tensile specimens were tested immediately or after an interval. Storage caused a considerable rise in the elongation and reduction of area. On storing the test pieces the improvement set in after a few days or weeks, but on storing the entire rails months were required before the properties became constant. Reheating at 200 deg. to 600 deg. Cent. produced an immediate improvement in elongation and reduction of area. The results were quite different from those which would be produced by a precipitation phenomenon. The yield point and tensile strength were unaffected, and yet the more complete the change the greater was the rise in elongation and reduction of area. The change was shown to be unconnected with rolling or cooling stresses, and the cause was supposed to lie in the equalisation of structural stresses, or stresses existing between the different elements of the microstructure.

The question has again been raised in the recently reported investigation of C. Drescher and R. Schäfer† on "Hydrogen as a Cause of Low Elongation and Reduction of Area in Steel."

They observed, in similar casts of low carbon steel, strikingly different values of elongation and reduction of area which could not be explained on the basis of segregation, inclusions, grain size, or the amount of work done on the steel in rolling. They therefore made an exhaustive study of a basic open-hearth steel (carbon, 0·16 per cent.), which for no clear reason showed unsatisfactory elongation and reduction of area in the rolled condition (Table I). The changes brought about by different conditions of annealing at 900 deg. Cent. might be dependent on grain size characteristics, but this does not apply to the improvement effected by treatment at 500 deg. Cent. Even heating at 100 deg. Cent. produced an improvement in a smaller bar of the same steel (Table II). The improvement was in all instances

TABLE I.—*Tensile Tests on Steel Bars, 36 mm. Diameter, in the Rolled Condition and After Different Heat Treatments.*

Specimens, 20 mm. diameter and 100 mm. long, of steel of the following percentage composition :—C, 0·16 ; Si, 0·9 ; Mn, 0·99 ; P, 0·042 ; S, 0·04 ; Cu, 0·38.

Condition.	Yield point, kg./mm.²	Max. load, kg./mm.²	Yield ratio, per cent.	Elongation, per cent.	Reduction of area, per cent.
As rolled	39	56	69	22	39
Heated for 30 min. at 900 deg. Cent. and cooled in air	41	56	73	25	45
Heated for 6 hours with the furnace held at 900 deg. Cent. 30 min., and cooled in air	39	56	70	29	52
Heated with furnace held at 900 deg. Cent. for 30 min., and cooled in the furnace	38	55	70	33	63
Heated 2 hours at 500 deg.	39	55	70	23	49
Heated 4 hours at 500 deg.	39	55	70	25	51
Heated 8 hours at 500 deg.	39	56	69	30	59

TABLE II.—*Tensile Tests on Steel Bars, 20 mm. Diameter, of Steel in Table I, in the Rolled Condition and After Treatment at 100 deg. Cent.*

Condition.	Yield point, kg./mm.²	Max. load, kg./mm.²	Yield ratio, per cent.	Elongation, per cent.	Reduction of area, per cent.
As rolled	43	59	73	24	40
Heated 8 hours at 100 deg.	43	59	73	28	56
Heated 12 hours at 100 deg.	43	57	75	30	64

TABLE III.—*Tensile Tests on Steel Bars, 40 mm. Diameter, in the Rolled Condition and After Treatment at 300 deg. Cent.*

Percentage composition of steel :—C, 0·09 ; Si, 0·1 ; Mn, 0·4 ; P, 0·021 ; S, 0·036.

Condition.	Yield point, kg./mm.²	Max. load, kg./mm.²	Elongation, per cent.	Reduction of area, per cent.
As rolled	21·6	38·2	31·0	52·2
Heated 12 hours at 300 deg.	22·3	37·5	37·5	69·2

* *Mitt. K.W. Inst. Eisenforschg.*, 1935, 17, 89 ; *Stahl und Eisen*, 1935, 55, 914.

† *Archiv für das Eisenhüttenwesen*, January, 1936, 1935/36, 9, 327.

accompanied by no change in the tensile strength, but by a remarkable alteration in the appearance of the fracture. While the specimens in the rolled condition gave a spongy speckled fracture, those which had been held for a sufficiently long time at a given temperature broke with a tough cup-and-core fracture, in which no spots could be seen.

A discussion of possible causes of this behaviour led the authors to the assumption that the presence of hydrogen under pressure in the steel played a part in bringing about the low elongation and reduction of area of the rolled bar.

Steel from a cast which had always given a good fracture with high elongation and reduction of area was remelted in a magnesite crucible and hydrogen injected for seven minutes. The melt was quickly cast into two thin-walled ingot moulds and the head of the ingot immediately quenched to prevent escape of the gas. The ingots were forged to 30 mm. diameter, and tensile test pieces taken, one being tested without heat treatment and the other heated for twenty-four hours at 100 deg. Cent. The untreated specimen gave a speckled fracture and very low values of elongation and reduction of area, while that which had been reheated at 100 deg. Cent showed a normal fracture free from specks.

The steel first employed (Table I) was rich in silicon, and, to ascertain whether the hydrogen was derived from the ferro-silicon used, a low silicon steel in which elongation and reduction of area were not specially high, was next investigated. Its behaviour (Table III) showed that ferro-silicon could not be regarded as the exclusive source of the hydrogen, but rather that the cause was already present in both steels.

It was found that unkilled steel responds to the treatment outlined above only in large sections. This was attributed to the low hydrogen content of unkilled steel, in which during solidification a lively

TABLE IV.—*Tensile Tests Before and After Treatment of Rolled Bars of Different Sizes.*
Steels containing 0·09 to 0·29 per cent. carbon.

Diameter.	Condition.	Yield point, kg./mm.²	Max. load, kg./mm.²	Elongation, per cent.	Reduction of area, per cent.
18 mm.	As rolled	43	59	27	40
	Heated 8 hours at 100 deg.	43	57	33	64
22 mm.	As rolled	45	67	18	29
	Heated 12 hours at 100 deg.	47	64	24	66
55 mm.	As rolled	41	59	22	31
	Heated 20 hours at 300 deg.	41	60	30	57
75 mm.	As rolled	28	42	27	35
	Heated 24 hours at 300 deg.	28	42	36	65
90 mm.	As rolled	22	38	31	45
	Heated 20 hours at 350 deg.	22	38	37	69
120 mm.	As rolled	36	57	24	34
	Heated 24 hours at 400 deg.	35	56	30	58

TABLE V.—*Tensile Tests on Slightly Worked Steel having Percentage Composition :*
C, 0·16 ; Si, 0·98 ; Mn, 1·14 ; P, 0·032 ; S, 0·04.

Condition.	Yield point, kg./mm.²	Max. load, kg./mm.²	Yield ratio, per cent.	Elongation, per cent.	Reduction of area, per cent.
As rolled	37	59	63	18	22
Heated 12 hours at 100 deg.	38	59	65	28	59

TABLE VI.—*Tensile Tests on Cast Steel of Different Cross Sections.*
(C, 0·18 ; Si, 1·09 ; Mn, 1·10 per cent.)

Size for low-temperature treatment.	Condition.	Yield point, kg./mm.²	Max. load, kg./mm.²	Elongation, per cent.	Reduction of area, per cent.
—	As cast	—	54	7	8
20 mm.	As cast, heated 60 hours at 100 deg.	—	58	24	36
—	Normalised	41	61	18	18
20 mm.	Normalised, heated 48 hours at 100 deg.	40	61	28	49
—	As cast	—	54	8	10
50 mm.	As cast, heated 48 hours at 300 deg.	—	58	25	32
—	Normalised	—	61	18	18
—	As cast	—	52	5	8
105 mm.	As cast, heated 60 hours at 400 deg.	—	58	22	30
—	Normalised	38	60	18	20
105 mm.	Normalised, heated 60 hours at 400 deg.	38	58	25	41

evolution of carbon monoxide occurs, carrying a large part of the hydrogen with it. The hydrogen is so low as to be detrimental only in large sections from which escape by diffusion is difficult. To test this point specimens of killed steels with different sizes of cross section were heated for long times at temperatures varying from 100 deg. to 400 deg. Cent. A few selected results from a large number given are contained in Table IV.

With increasing thickness of section, it was found necessary to apply a longer time or a higher temperature of treatment to achieve the same result. The necessary time and temperature of heating bear a linear relationship to the thickness of the section of rolled material which may be expressed by saying that for a duration of heating of twenty-four hours the effective temperature rises with thickness of section from 100 deg. Cent. for 20 mm. to about 480 deg. Cent. for 120 mm. Similarly, if the same temperature 350 deg. Cent. is used for all sections, the treatment is effective in six hours in steel 20 mm. thick and requires thirty-six hours in steel 120 mm. thick.

To determine the effect on material which had received little work, an ingot of 0·16 per cent. carbon steel, 220 mm. square, was rolled to 130 mm. square, and two test pieces, 20 mm. in diameter, were taken for tensile tests, one in the rolled condition and the other after twelve hours' heating at 100 deg. Cent. (Table V). Here, again, the yield point and maximum load were unaltered, but the values of elongation and reduction of area were clearly raised, while the brittle and speckled fracture of the rolled bar was transformed by heating at 100 deg. Cent. to a tough cup-and-cone fracture. Results of further experiments on steel in the cast condition are indicated in Table VI. The elongation and reduction of area of the cast steel are raised and almost attain the values for the rolled condition, provided that the time and temperature of heating are sufficient. It may be noted that the cast steel requires a longer time at a given temperature than the rolled steel. The effective temperature rises with thickness from 100 deg. Cent. for 20 mm. to about 480 deg. Cent. for 120 mm. for a duration of heating of sixty hours (instead of the twenty-four hours which were sufficient for the rolled steel), and if the same temperature, 350 deg. Cent., is used for all sections the treatment is effective in thirty hours in cast steel, 20 mm. thick, but requires sixty-five hours in cast steel 120 mm. thick. The time required was thus three or four times as long as for the rolled steel. An attempt to reduce the necessary time by previous normalising was not successful. The behaviour of the cast steel differed from that of the rolled steel in that a rise in tensile strength also accompanied the improvement in elongation and reduction of area. This, however, was probably due to the elimination of premature failure of the test pieces which had occurred in the somewhat brittle cast condition.

In this research a phenomenon which has long been known or suspected, viz., that elongation and reduction of area may be improved by prolonged heating at raised temperatures, was fully explored for mild steel (though further information is required on the behaviour of the steel in bend tests and notched bar impact tests), and a suggestion, based on Bennek, Schenck, and Müller's theory of the formation of flakes in alloy steels,‡ was put forward to explain the results. First, it may be emphasised that Drescher and Schäfer do not identify their speckled fractures with flakes. They find the effect in mild steels with under 0·3 per cent. of carbon. The fracture is affected by *flecken*, not *flocken*. Flakes are present in a steel before testing; the specks are formed in the fracture of the tensile test piece. Secondly, it seems quite clear that heating at 100 deg. Cent. effects a genuine improvement in many steels. One confirmatory series of tests put forward by Schottky was made on 0·2 to 0·3 per cent. carbon steel, melted under hydrogen, killed with aluminium, cast, quenched and forged to 40 mm. diameter. Comparative tests before and after treatment for twenty-four hours at 100 deg. Cent. gave the following results.

	Untreated.	Heated 24 hours at 100 deg. Cent.
Elongation, per cent.	10·9 to 13·5	18·9 to 21·8
Reduction of area, per cent.	12 to 23	36 to 50

Yield point and maximum load remained unchanged. Although Drescher reported that no change in elongation and reduction of area was produced by treatment at 100 deg. Cent. of steels with more than 0·4 per cent. of carbon, the results appear to be linked up very closely with Körber and Mehovar's work on rail steels to which reference has already been made. According to their explanation, the improvement of ductility was due to the release of structural stresses, but they did not specify any cause for the stresses themselves. Drescher and Schäfer's explanation may be quite consistent with theirs, but it goes further, and assigns as a cause for the stresses the pressure of hydrogen falling out of solution at a rate faster than it can diffuse away through the steel. After hearing of Drescher and Schäfer's results, Körber and Mehovar made actual determinations of hydrogen content on a number of their specimens, with concordant results, the averages of which are shown in Table VII. These results offer strong support to Drescher and Schäfer's theory though they leave unexplained the fact that Drescher and Schäfer found no effect in steels with more than 0·4 per cent. carbon. This discrepancy will, however, most probably yield to further investigation, and meanwhile the paper of Drescher and Schäfer forms a valuable record of the important improvement which it may be possible to effect in some steels by means of a treatment which can be very cheaply and easily carried out.

TABLE VII.—*Tensile Tests and Hydrogen Content of Rail Steels in Different Conditions.*

Condition.	Max. load, kg./mm.²	Elongation, per cent.	Reduction of area, per cent.	Hydrogen content, c.c. per 100 g.
Tested immediately after rolling	88·0	9·1	10	2·13
Specimens heated 1 hour at 200 deg.	88·7	13·1	20	0·10
Specimens stored seven weeks	89·0	12·9	21	0·10
Rail stored seven weeks	86·0	12·3	19	0·36

‡ *Stahl und Eisen*, 1935, 55, 321.

Internal Fissures in Rails.

By J. C. W. HUMFREY, O.B.E., B.A., M.Sc., M. Eng.

No. I.

HISTORICAL.

THE existence of a particular species of fine internal crack in steel forgings was brought to the attention of metallurgists during the war, when their presence was discovered in certain gun tube and aeroplane engine forgings, particularly those of nickel-chrome steel. The gun forgings were tested by tensile and bend test pieces cut tangentially from a transverse slice. In certain cases test pieces fractured at an abnormally low load, practically without elongation or reduction of area, and the fractured surfaces showed all or part of a circular patch up to ½in. diameter of totally different texture to the remainder or to that of a normal test piece. While it was evident that these circular patches represented areas of no adhesion, no visible defect could be seen on the ground surface of the test pieces before breaking. Examination of polished cross sections from the gun tubes again showed no apparent defects, but on deeply etching the surfaces numerous small radial cracks were revealed lying usually nearer to the centre than the exterior. No very definite relation could be found between the path of the cracks and either the macro or micro-structure of the steel, although their branching form suggested an intercrystalline path.

No explanation as to the cause of the cracks was then available, although for some years after the war numerous workers—especially in America—put forward various theories as to conditions of steel melting, casting, and forging which produced a tendency to their formation.

In 1925 Hultgren presented a paper to the Iron and Steel Institute[1] dealing with the existence of such cracks in chrome steel forgings used for ball and roller bearings and clearly demonstrated how in such material :

(*a*) The cracks were undoubtedly formed during air cooling after the final forging ;

(*b*) Their formation could be prevented if this cooling was sufficiently retarded from a temperature of 760 deg. Cent.

It was not denied that cracks could possibly have formed during previous coolings, *e.g.*, during cooling of the ingot, but it was pointed out that practically all such cracks would be welded up during subsequent forging, and that those found when forging was completed were formed during the final cooling stage. On the other hand, it could be gathered from Hultgren's investigation that if internal cracking was prevented by slow cooling after the final forging, then the forgings could subsequently be annealed, reheated, and quenched without fresh cracks appearing.

Hultgren concluded his paper by drawing attention to the similarity of the defects he had found in his chrome steel forgings with those which had previously been found in certain high carbon rails in America, and which certain investigators at that time believed to be the original cause of transverse fissure rail fractures.

TRANSVERSE FISSURE RAIL FAILURES.

The serious nature of rail failures due to transverse fissures which have occurred in U.S.A. may be gathered from the following statistics which are taken from the Report of the Rail Committee of the American Railway Engineering Association for January, 1934. Up to and including December 3rd, 1932, 65,281 transverse fissure failures had been reported on American railways, representing an average of 6·14 failures per 100 miles of track per annum. Since 1928 a proportion (amounting to 36 per cent. in 1931 and 1932) of these failures were detected by means of the Sperry car, which enabled the faulty rails to be removed from the track before an actual fracture occurred, but in each of these two years about 4700 failures took place which had not been previously detected.

The true transverse fissure failure is due to a fatigue crack starting in the interior of the rail head and gradually spreading upwards and outwards until the section is so weakened that a sudden complete transverse break may occur under a passing wheel load. In many cases where a rail has so failed, subsequent examination has revealed the presence of a number of such internal fatigue fractures at various positions along its length and in various stages of development. Rails may, of course, break transversely in track owing to many causes other than transverse fissures, such as segregation and other ingot defects, or to the spread of small cracks formed on the running surface due to skidding wheels. Transverse fissure failures have, however, certain definite characteristics which make them very easily distinguished from failures due to other causes.

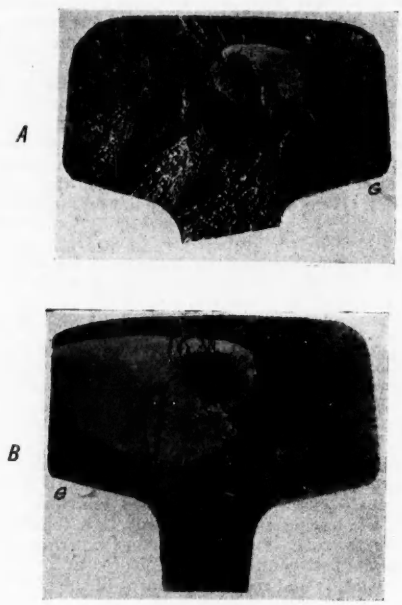

FIG. 1

Typical examples of a transverse fissure failure are shown in Fig. 1,[2] and attention should be drawn to the areas X of a texture different from either the

[1] *Journal* of the I. and S. Inst., 1925, No. I, page 113.

[2] Reproduced from a paper by Rawdon, "The Presence of Internal Fractures in Steel Rails, &c.," "Trans.," Faraday Society, 1921.

smooth fatigue fracture surrounding them, or the sudden fracture through the remainder of the section. The significance of this area has led to considerable controversy, but it is now generally agreed that it represents an original internal crack or "shatter" which was present in the rail before it was laid in the track, and which was the definite cause and origin of the failure. Definite experimental evidence of this fact has recently been obtained by a special committee of the American Railway Engineering Association, which has carried out a very extensive investigation of the subject at the University of Illinois.[3]

The internal cracks found in a rail before laying in track may run either transversely or longitudinally. Both types may lead to failure of the rail in track, the former by the dangerous transverse fissure type which as been described above; the latter to a rather less dangerous type known as a split head. Split head failures due to internal fissures usually show up very clearly before any portion of the rail actually breaks away, by the rail head spreading out sideways as the internal crack joining up the original fissures gradually opens out. Such rails can therefore usually

FIG. 2

be removed from the track before serious failure occurs. Compound fractures originating at both a longitudinal and transverse internal crack are also met with, and cases are known of failures due to internal cracks lying parallel to the tread and resulting in a breaking away of a horizontal layer from the top of the head.

The presence of shatter cracks in the interior of the head of certain rails as delivered from the mill and before laying in the track was first discovered by Waring and Hofamman in 1919, by deeply etching sections from the head. The fact that the defects found in this manner were actual pre-existing cracks, and were not some other form of local weakness which developed into a crack during etching, was questioned by several speakers in the discussion on this paper; but the matter was later conclusively settled by Rawdon, who revealed the presence of actual cracks, without etching, by magnetising the sections and immersing them in kerosene containing fine iron dust in suspension. The discontinuity in the magnetic flux at the cracks caused the iron dust to adhere in such a manner that they were clearly defined. Cracks found in this way could be subsequently examined microscopically, without etching, or broken open and their true nature definitely established.

Howard, who as Engineer Physicist to the American Interstate Commerce Commission had investigated many cases of accidents caused by transverse fissure failures, accepted the existence of shatter cracks, and advanced a theory that their formation was due to "hydraulic" tensile stresses set up in the interior of the head during cooling. He was never convinced, however, that shatter cracks were the necessary origin of transverse fissures and contended that stresses due to wheel pressure, added to internal stresses present in the rails when laid, might cause a transverse fissure failure in rails free from shatter. Subsequent investigation, particularly that recently published by the University of Illinois, has, however, failed to confirm this view.

The elimination of transverse fissure failures has thus been reduced to that of the prevention of shatter cracks in rails before they are laid in track, and we can therefore proceed to consider the conditions under which such cracks are formed and methods which can be adopted to eliminate them.

THE FORMATION OF INTERNAL COOLING CRACKS.

Hultgren showed how the nature of the "flakes" found in his chromium steel forgings and especially their almost perfect circular area when broken open necessarily limited their formation to cooling after final forging, and similar observations made on the shatter cracks found in certain rails again give definite evidence of their having been formed after the last pass through the finishing rolls.

A horizontal section through the head of a badly shattered rail after deep etching is shown in Fig. 2. Fig. 3 shows a fracture obtained by original detection of a crack with the iron dust method and subsequently breaking through by bending. The very approxi-

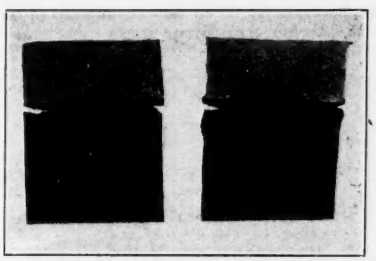

FIG. 3

mately circular area of the original crack should be noted, and the difference of texture of the fracture in the area of the crack to that in the surrounding metal.

If the similar nature of such cracks in rail steel to those found in alloy steel forgings be granted, then there is no doubt that their formation could be prevented by similar means to those found by Hultgren, *i.e.*, slow cooling of the rails from 750 deg. Cent. after leaving the last pass through the rolls, but such a procedure would involve some very undesirable features. A rail is subjected in the track to severe battering, loading, and abrasion on its running face, which necessitate a maximum hardness (and especially high elastic limit) consistent with adequate toughness to withstand shock loading as a girder. Very slow cooling through the critical range from 750 deg. Cent. would, however, tend to reduce both hardness and elastic limit to a minimum and therefore tend to give poor wearing qualities.

Again, careful investigation of many consecutive heats of rail steel in America has shown that only a

[3] See THE METALLURGIST, February, 1936, page 107.

proportion of such heats—and in some cases even only a proportion of the ingots from a heat—tend to form shatter cracks during normal cooling, and there is obviously some condition or conditions which occur during melting, casting, or rolling which produce this tendency and which, if they could be avoided, would give shatter-free rails. While some recent investigations[4] have indicated that the presence of hydrogen in steel may produce a marked tendency to shatter, it does not appear that even the most careful records of a heat from steel furnace to hot bed can give definite evidence as to whether the rails made from it are shattered or sound. Even if means of avoiding or minimising the tendency to shatter during the manufacture of the steel are finally determined, it may yet be more economical and safer to eliminate it at a later stage as described below.

THE SANDBERG OVEN PROCESS.

A process patented by Messrs. Sandberg in 1928 is now in extensive operation in many countries, by means of which shatter in rails can be completely eliminated, even in highly susceptible heats, without any detriment to their wearing qualities, and which, in fact, allows of accelerated cooling of the rails through the critical range so that their wearing qualities can be very greatly improved.

The theory of the process was based upon a study of the variation in the strength and ductility of steel in cooling down from the critical temperature to that of the atmosphere. Tensile tests on steel of various analyses were published by Dupuy in 1921[5] and his results are plotted in Fig. 4. Some results of slow

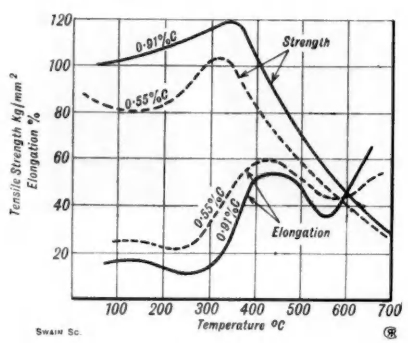

FIG. 4

notched bar tests on American rail steel (on a machine of the author's design) obtained in Messrs. Sandberg's laboratory during their investigation into the subject of internal fissures are shown in Fig. 5. Both series of tests show the same general characteristics, *i.e.*, a very rapid rise in strength after cooling through the critical range to a maximum at about 350 deg. Cent. and a subsequent fall to a minimum between 100 deg. and 200 deg. Cent. There are also well-marked minima of ductility between 450 deg. and 550 deg. Cent. and at 200 deg. to 250 deg. Cent. From such experiments it appeared that the important range of cooling as regards the formation of internal fissures was that between 550 deg. and 350 deg. Cent.—that is, when the strength is very rapidly increasing with falling temperature. Rails leave the mill with a probable considerable temperature difference between the interior and exterior portions of the head, owing to the cooling action of the rolls on the surface and the transformation of work into heat in the interior as the section is gradually formed and reduced. During the early stages of cooling the less massive parts of the section and the exterior portion of the head cool the fastest, but since all parts finally arrive at a uniform temperature, *i.e.*, that of the atmosphere, at some later stage, this condition must be reversed, and the interior be cooling more rapidly than the exterior. Such a condition may imply tensile stresses in the interior and if these occur when, owing to temperature differences, the interior is weaker and

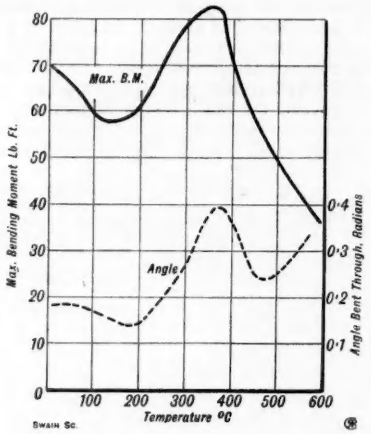

FIG. 5

less ductile than the exterior—such as may be the case between 550 deg. and 350 deg. Cent.—then conditions are such that the formation of internal fissures may occur.

The probability of this range of cooling, 550 deg. to 350 deg. Cent., being that of fissure formation was further indicated by microscopical examination of the fractures of the notched bar bend test pieces. Microscopical examination of typical internal fissures found in unused rails indicates that their path is round, and not through the pearlite grains. A similar type of fracture was found in the test pieces broken in the range of 400 deg. to 550 deg. Cent. in contrast to fractures through the pearlite grains which were typical of test pieces broken at lower temperatures. In the 400 deg. to 550 deg. Cent. range also small isolated cracks could frequently be found, apart from the main fracture showing a characteristic path round the grains.

(To be continued.)

[4] See THE METALLURGIST, February 28th, 1936, page 99.

[5] *Journal* of the I. and S. Inst., 1921, No. 2, page 91.

The Influence of Alloy Additions on the Scaling of Steel.

A COMPREHENSIVE study of the influence of special elements on the scaling of steel has been made by E. Scheil and K. Kiwit,* who have carried out

* *Archiv für das Eisenhüttenwesen*, February, 1936, 1935–36, 9, 405.

chemical, microscopical, and X-ray examination of the scale, as well as gain of weight experiments, on low-carbon steels prepared from Armco iron, with additions of aluminium (to 16 per cent.), beryllium (4 per cent.), chromium (to 30 per cent.), copper (to 2·2 per cent.), manganese (to 4 per cent.), molyb-

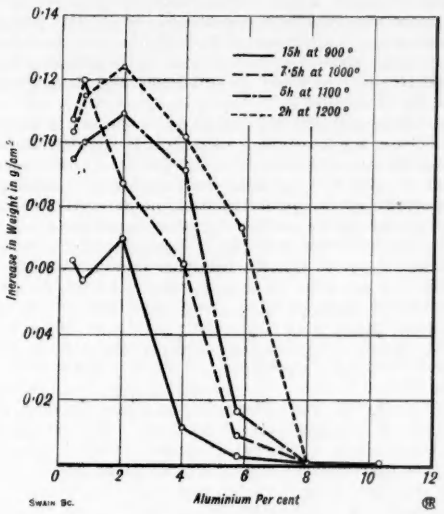

FIG. 1

denum (to 10 per cent.), nickel (to 2·2 per cent.), phosphorus (to 0·3 per cent.), platinum (1 per cent.), sulphur (to 0·16 per cent.), silicon (to 11·5 per cent.), titanium (to 6 per cent.), vanadium (4 per cent.), tungsten (4 per cent.). The steels, made, with two

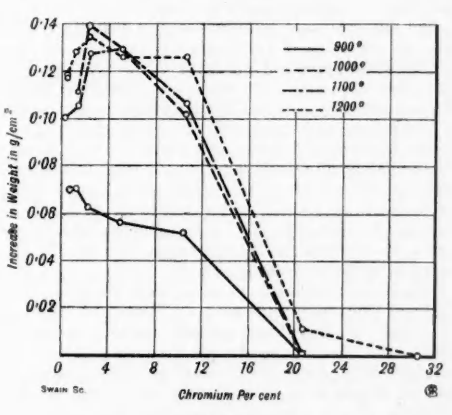

FIG. 2

exceptions, in 3 kilos. ingots, were forged down and scaling tests carried out on cylinders 50 mm. long by 10 mm. diameter, polished on 00 emery paper. During the scaling process at 900 deg., 1000 deg., 1100 deg., and 1200 deg. Cent., the specimen was suspended from a hook attached to a loop holding a small quartz crucible arranged so as to catch the scale which fell off. The individual layers in the scale were so sharply defined from one another that they could readily be separated for chemical study, and it was found that in all cases the alloy elements, whether "noble" or "base," became concentrated at the iron/scale boundary. Protective layers were formed by additions of aluminium, chromium, and silicon, which markedly raise the resistance to scaling (see

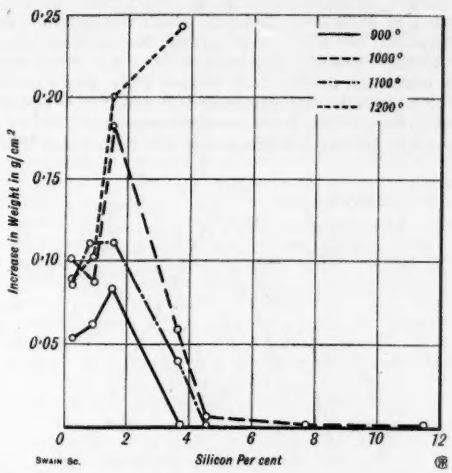

FIG. 3

Figs. 1, 2, and 3), as well as by titanium, beryllium, and vanadium, which increase it only slightly (see Table). The influence of additions of manganese,

TABLE.—*Increase in Weight of Some Experimental Steels after Scaling at Different Temperatures.**

Added element and percentage.	Increase in weight in g./cm.² after scaling for			
	15 hr. at 900 deg.	7½ hr. at 1000 deg.	5 hr. at 1100 deg.	2 hr. at 1200 deg.
Iron	0·07	0·11	0·10	0·12
Be 4	0·06	0·09	0·09	0·11
Cu 2	0·07	0·08	0·08	0·10
Mn 4	0·08	0·14	0·10	0·12
Ni 2	0·07	0·13	0·11	0·10
P 0·03.. ..	0·05	0·19	0·12	0·12
Pt 1	—	—	0·17	—
S 0·016 ..	0·07	0·10	0·12	0·11
Ti 4	—	0·04	0·07	—
V 4	0·08	0·09	0·10	—
W 4	0·06	0·09	0·08	0·12

* The results, given in the original to four places of decimals, are here reproduced to two places only.

sulphur, boron, molybdenum, tungsten, phosphorus, arsenic, and antimony was also studied, and here again in every case except manganese, enrichment of the added metal was found at the iron/scale boundary layer. The metals more "noble" than iron (of which copper, nickel, and platinum were studied) did not lead to an improved resistance to scaling; in fact, the steel containing platinum was strongly attacked. The behaviour of the nickel-chromium-iron alloys at 1000 deg. Cent. was investigated, and regions of relative resistance to scaling mapped out (Fig. 4). In some instances also the composition of the oxide layers was determined, and is indicated in the diagram. None of the steels resistant to scaling was of low alloy content. There are a few general

limitations which govern the behaviour of steels on scaling. For example, the steel itself must have a melting point well above the scaling temperature. This self-evident condition, however, restricts the useful temperature range of certain steels of high alloy content. Then the protective layer itself must have a high melting point, since after this is exceeded the protection to scaling ceases. This is shown by the example of silicon (Fig. 3), in which protection has failed at 1200 deg. Cent., at which temperature the eutectic $(FeO)_2SiO_2+SiO_2$ melts. No element which markedly lowers the melting point of the oxide must be contained in the steel, neither must any element which produces any appreciable amount of a volatile compound. The latter disadvantage has been held to apply to molybdenum steels, but Scheil and Kiwit

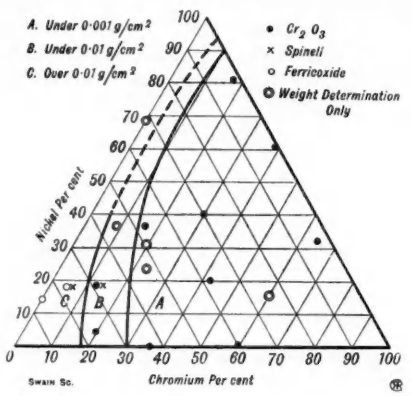

FIG. 4—*Effect of Nickel and Chromium Content on Scaling of the Steels Studied after 7½ hours' Heating at 1000 deg. Cent.*

find that although molybdenum is not a desirable constituent for scale-resisting steels, there is no marked volatilisation of the scale of an iron-molybdenum alloy even at 1200 deg. Cent. This they attributed to the formation of an infusible and non-volatile scale $FeO\,MoO_3$. Bischof observed marked gas evolution from the scale of chromium-molybdenum steels, but the difference is to be explained by the effect of chromium, since it is possible that chromium hinders the formation of the infusible compound $FeO\,MoO_3$, and so allows the volatile MoO_3 to be formed.

The observed phenomena, and particularly the distribution of the oxide of the added metal in the scale, can only be explained on the assumption that during scaling it is the iron atoms which migrate and not the oxygen atoms. The enrichment of the alloy element at the iron/scale boundary suggests that the iron, in preference to the other elements, diffuses outwards through the boundary layer and contributes to the building up of FeO. In order that a protective layer of scale may be formed, the new constituent due to the added element must have a low diffusion speed and solubility in ferrous oxide and must give rise to a continuous adherent film on the surface, hindering the outward diffusion of iron.

This extensive series of tests confirms the opinion, which might be derived from previous work, that chromium and aluminium are the only two elements which confer a pronounced resistance to scaling on steels, and that the presence of one or other of them is desirable in all alloys in which this quality is desired.

Cadmium-Base Bearing Metals.

MOST alloys for bearings consist of soft metals, such as tin or lead, with some addition which results in the formation of hard crystals embedded in the mass of softer metal. These hard crystals serve to support the load and resist wear, while the matrix wears away and leaves pockets which hold the lubricant, and, moreover, on account of its softness, it is able to yield to slight irregularities and shape itself closely to the shaft.

Certain cadmium alloys have the necessary duplex structure, and, although cadmium is slightly harder than tin and lead, it is still a comparatively soft metal, and so forms a suitable base metal for bearings, especially as its higher melting point gives it some advantage under high-temperature conditions.

Cadmium-base bearing metals usually contain either copper or nickel with other metals. Cadmium with 1·5 per cent. of copper and 1 per cent. of magnesium is said to have given excellent service for general plant and engineering work in Australia.* The silver-copper-cadmium bearing alloys have been developed by several automobile manufacturers in the United States. The General Motors Corporation† has patented an alloy for bearings having the following composition:—Cadmium, 95 per cent.; silver, 0·05 to 5 per cent., together with a hardening metal (particularly copper or nickel or both), 0·25 to 1·0 per cent.; while zinc may be present in amounts up to 0·25 per cent. It is essential that the highly plastic white metals should adhere strongly to the journal or liner in order to prevent their lateral flow under compressive load which might result in the alloy being squeezed out of the bearing. The cadmium alloys have the advantage of greater compressive strength and can be made to adhere strongly to steel. The bonding alloy for uniting the above composition to steel consists of 80 to 95 per cent. cadmium with the remainder zinc.

The silver-copper-cadmium alloys usually contain over 98 per cent. of cadmium with over 1 per cent. of copper and a small quantity of silver. Mr. L. C. Blomstrom‡ claims that silver improves the fluidity of the liquid alloy, makes the cadmium less susceptible to oxidation and improves the structure and properties of the final product, but states that the presence of copper permits considerable reduction in the silver without loss of hardness. This alloy has a melting point of 315 deg. Cent. and a tensile strength which is considerably greater than that of a tin-base bearing metal at atmospheric temperature, and nearly three times as great at 200 deg. Cent.

The cadmium-nickel bearing alloys consist essentially of pure cadmium with 1·3 per cent. of nickel, to which is sometimes added copper, magnesium, silver, and zinc. The duplex structure of the alloy consists of a hard nickel-cadmium compound embedded in a matrix of relatively soft cadmium-nickel eutectic. Tests carried out on this alloy have been described by Mr. A. J. Phillips.§ The coefficient of friction was found to be lower and the fatigue strength higher than that of a tin-base bearing alloy. "Initial softening" is said to occur at 318 deg. Cent. This

* "Proc.," Australasian Inst. Min. Met., 1934, page 201; *Commonwealth Engineer*, 1934, 22, 81.

† Patent Specification No. 436,633.

‡ *Machinist* (European Edn.), January 18th, 1936, 768.

§ *Machinist* (European Edn.), December 21st, 1935, 709.

temperature must, however, refer to the melting point, since that of pure cadmium is only a few degrees higher. Cadmium-nickel bearings, it is claimed, have now been in satisfactory service under severe conditions for three years, and have established their position in fields of application involving high temperatures and heavy loading.

The increased use of cadmium-rich bearing alloys during 1935 led to a shortage of the cadmium required for its previously well-established application as a protective coating on steel. Since cadmium is a by-product metal and a limit is set by the output of zinc to the supplies available, this shortage and the consequent rise in price has become a serious embarrassment to the electro-plating industry. When other possible uses of cadmium are taken into account, *e.g.*, for solders, fusible alloys, type metal, and as an addition to copper for electrical purposes, it seems probable that cadmium will be one of the few metals for which the demand during the next few years is likely to exceed the supply.

Metals under High Pressures.

The commercial utilisation of high pressures has revolutionised some branches of chemistry and many physical effects in which metals are involved are profoundly influenced by pressure. Among these may be instanced the action, under high pressure, of mercury or of hydrogen in producing intercrystalline failure of steel at atmospheric temperature. Many of the low melting point metals act similarly under high pressure. Sodium will cause rupture of a steel cylinder and so also will tin, bismuth, and other metals. High pressures also will aid the rapid surface attack of steel by sulphur. The experimental study of the behaviour of metals under high stresses has hitherto been confined almost exclusively to their behaviour under high hydrostatic pressure. Dr. P. W. Bridgman, of Harvard University, who is well known for his pioneer researches on the effects produced by high hydrostatic pressure, has recently published an account of his work on the "Effects of High Shearing Stress Combined with High Hydrostatic Pressure."*

If a metal is squeezed between two cylinders of hardened steel with accurately ground ends, it will flow out laterally until the thickness of the disc is reduced to such a value that the friction against the steel near the outer edge of the disc is sufficient to balance the hydrostatic pressure on the central parts of the disc. If, when this equilibrium thickness has been reached for a given pressure, a couple is applied to one cylinder while the other is held stationary, a shearing stress is set up in the disc, the maximum magnitude of which is determined by the force necessary to produce uniform rotation of one cylinder in relation to the other. This, in principle, was the method employed by Bridgman. In practice the arrangement was doubled. Two pistons, acting in opposite directions, were separated by an anvil between which and each piston a disc of the material under test was held. The two pistons were compressed together with any desired force with a hydraulic press and the force required to rotate the anvil was measured by means of a simple strain gauge on the handle of a rotating wrench. The mean pressure exerted by the pistons was carried up to a maximum of 50,000 kilos. per square centimetre. The maximum force required to produce rotation varied greatly from substance to substance, the greatest value reached for any substance being 18,000 kilos. per square centimetre. The actual thickness of the discs after exposure to a mean pressure of 50,000 kilos. per square centimetre was only a few thousandths of an inch.

* *Physical Review*, November 15th, 1935, 48, 825.

Effect on Chemical Reactions.

Before the experiments on the metals are described it may be of interest to refer briefly to certain effects noted in non-metallic materials, and to the influence of the conditions of experiment on some chemical reactions involving metals. Thus, paper was transformed into a translucent horn-like mass, though hydrostatic pressure alone had no such effect; the rotation was also necessary. Celluloid detonated violently. So also did lead peroxide, leaving a residue of metallic lead. With litharge between anvil and pistons the anvil could be rotated quietly under a compression of 50,000 kilos. per square centimetre, but when the pressure was released the litharge was found to have decomposed completely, leaving a film of metallic lead. The question of the temperature rise in the discs is of importance, but Bridgman calculates that the heat due to friction would be so quickly dissipated that the rise in temperature in the disc for this reason would not exceed 37 deg. Cent. Many salts were found to detonate, *e.g.*, potassium permanganate, ammonium dichromate (but not potassium dichromate), silver nitrate (but not mercuric nitrate), and, of course, a great number gave negative results.

Between copper and sulphur, combination occurred with detonation at a pressure of 20,000 kilos. per square centimetre (without rotation), the product being ordinary black copper sulphide. A mixture of ferric oxide and aluminium detonated at between 20,000 and 30,000 kilos. per square centimetre without rotation. In the following instances one specimen out of two was detonated by rotation at 50,000 kilos. per square centimetre, but the other could not be made to detonate by prolonged rotation at this pressure:—Potassium oxalate and aluminium, silica and magnesium, magnesia and silicon.

Effect on Metals.

The quantitative experiments on metals consisted of a determination of the force required to rotate the anvil as a function of the mean pressure. A complete set of readings was taken up to 50,000 kilos. per square centimetre, and back to zero in steps of 2000 kilos. per square centimetre. In the first stage the stresses are so low that there is no plastic deformation. The disc slips over the face of the steel piston. With increasing normal pressure the second stage begins in which seizing of one surface to the other occurs, usually with an increase in the coefficient of friction. When the whole surface has seized, the third stage is entered, in which shear is brought about by plastic flow distributed more or less uniformly throughout the entire body of the material. Only with the softer metals was there complete seizing to the steel. With all the metals there was usually a certain amount of welding at a small number of small spots, but the disc could almost always be pried away from the anvil with a razor blade as a coherent whole. The effect of cold working was observed in the fact that the curve with decreasing pressure often lay above that obtained with increasing pressure. In many of the harder metals, however, this work-

hardening was unimportant. The typical metallic crystal structures permitted plastic flow to take place quietly, while in non-metallic elements the shear occurred in violent jerks. This was so in the case of the typical non-metallic elements, boron, carbon, silicon, phosphorus, and arsenic (but not sulphur). The crystal structure is not completely determinative because bismuth and antimony both have the same structure as arsenic but shear quietly. There was no close correlation between the smoothness of the shearing and the numerical value of resistance to shear.

An important feature of the paper is the deduction which Bridgman makes with reference to the polymorphism of some of the metals. In a large number of instances the shearing stress when plotted against mean hydrostatic pressure gave a smooth curve, in others it gave discontinuities and sudden changes in direction which Bridgman attributed to the existence of transition pressures between allotropic modifications of the element. Such modifications might conceivably be more easily recognised in this way than by measurements of volume changes. On the other hand, it is clear that in the absence of confirmation

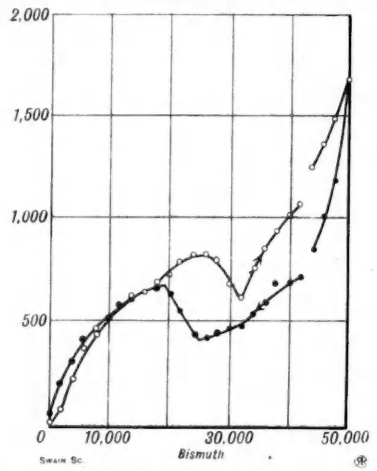

FIG. 1—*The Curve of Shearing Stress of Bismuth in kg/cm² (Ordinates) against Mean Hydrostatic Pressure in kg/cm² (Abscissæ). Open circles obtained with increasing pressure, solid circles with decreasing pressure. The breaks in the curves indicate polymorphic transitions.*

from other sources there can be no certainty that the breaks are due to polymorphism.

In a later paper, "Polymorphism, Principally of the Elements, up to 50,000 kg./cm.²,"† Bridgman has confirmed, by measurements of volume changes, the existence of transition pressures for some (but not all) of the elements which showed breaks in the shearing stress curves. The only common metal of which the polymorphism under pressure was confirmed was bismuth. The curve of shearing stress of bismuth against mean hydrostatic pressure, shown in Fig. 1, suggests polymorphic change at 20,000 to 30,000 kilos. per square centimetre and possibly also in the vicinity of 42,000 kilos. per square centimetre. Bridgman's later work confirms these changes, to represent which he gives the phase diagram Fig. 2.

The values of the maximum shearing stresses under high compression are of interest in themselves. Usually they were from three to twenty times the shearing strength at ordinary pressures. In the Table they are recorded in order of the atomic numbers of the elements. Those which showed breaks in the curve are omitted from this list. Among the more common elements omitted for this reason are Li, Ca, Sr, Ba, Zn, V, Mn, Sb, Bi, and Sn. The table shows a certain regularity in the values which rise to a maximum and fall away again several times, the successive maxima being at boron (group III of the

† *Physical Review*, December 1st, 1935, 48, 893.

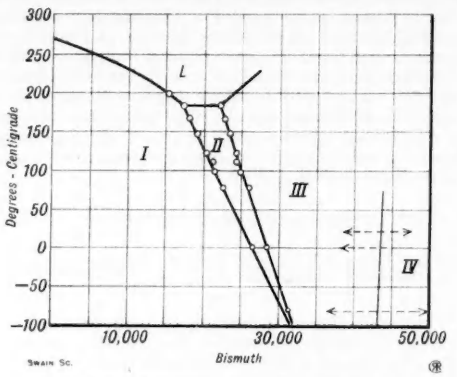

FIG. 2—*The Phase Diagram of Bismuth, Pressure in kg/cm² as Abscissa Against Temperature in deg. Cent. as Ordinate. The arrows on the transition line III-IV mark the limits of indifference.*

periodic system), silicon (group IV), titanium and vanadium (groups IV and V), arsenic (group V), molybdenum (group VI), tungsten or osmium (groups VI and VIII), since rhenium is perhaps a doubtful value. The elements showing maximum values are displaced across the periodic table as the atomic weight rises.

TABLE—*Shear Strength of Different Elements under 50,000 Kilos. per Square Centimetre Compression (Omitting those which Showed Breaks in the Curve).*

Element and atomic number.	Shearing strength, kg./cm.².	Element and atomic number.	Shearing strength, kg./cm.².
Be 4	7,400	Yt 39	4,300
B 5	18,000	Zr 40	4,400
C 6	10,000	Nb 41	9,000
Na 11	5,200	Mo 42	12,100
Al 13	3,200	Ru 44	9,600
Si 14	9,000	Rh 45	8,900
P 15	8,300	Pd 46	5,600
S 16	7,000	Ag 47	4,700
Ti 22	13,000	In 49	750
V 23	13,000*	Ta 73	11,500
Cr 24	12,300	W 74	12,700
Fe 26	10,600†	Re 75	7,500
Co 27	6,300	Os 76	17,400
Ni 28	8,700	Ir 77	8,400
Cu 29	4,700	Pt 78	5,800
Ge 32	5,700	Au 79	4,500
As 33	11,200	Pb 82	710
Se 34	5,500	U 92	8,900

* Curve not always smooth.

† At 42,000 kilos. per square centimetre, the maximum in this experiment.

Many of the elements which show breaks in the curves fit into the series given in the Table, the most marked exceptions being zinc and manganese. Regularity can hardly be expected, however, in

metals which, in all probability, have undergone a polymorphic change.

In this work Dr. Bridgman has carried the investigation of the properties of metals into a field which his achievements in high-pressure research make him most fitted to explore, and further results will be awaited with interest.

Seizing of Screw Threads.

The seizing of screw threads, as, for example, in a nut and bolt, is not necessarily similar in its origin to the seizing of a continuously running part, such as a shaft in its bearings. In the first instance, static effects are predominant, while in the second instance frictional effects, giving rise to local increase of temperature, may have an important influence. There must, of course, be some factors in common, such as the influence of surface finish and efficiency of lubrication. In general, the seizing of screwed connections, just like that of moving parts, is more likely to occur when the load carried is high than when the surface pressure between the two parts is small. Hence the seizing of screwed connections has been a serious trouble in high-pressure vessels used in chemical industry, especially if they are exposed to raised temperatures. Crushing, abrasion, or shearing at certain positions of the thread may occur during unloosening of the nut, and liability to this defect is increased after long exposure to temperatures which may rise to 500 deg. Cent. It was with a view to examining some of these factors that H. Vollbrecht undertook an investigation, the results of which have been published* under the title of "The Seizing of Screwed Connections at High Temperatures."

A detailed study has been made of the behaviour of a number of steels working under a variety of conditions involving temperatures up to 500 deg. Cent. Pairs were made up, in the manner shown in the table, from the following steels :—

(1) 0·11 per cent. carbon steel (German standard, "St 37")

(2) Chromium-molybdenum steel (C 0·40, Cr 1·5, Mo 0·30 per cent.).

(3) Chromium-nickel austenitic steel (C 0·22, Cr 20·0, Ni 8 per cent.).

(4) Chromium-aluminium-molybdenum nitriding steel (C 0·32, Cr 1·2, Al 0·93, Mo 0·34 per cent.).

Earlier experience had already shown the importance of the nature of the steels working in contact, and had suggested the importance of oxide films in inducing seizing, so the research described by Vollbrecht took account of differences in hardness of the two metals, the thickness of the oxide film on their surfaces, the time during which the screwed joints were heated, and the range of temperature to which they were exposed.

The mechanical properties are not given in the paper, but it is stated that the Brinell hardness number of steel 4 was three times that of steel 2 and 5·76 times that of the carbon steel. Measurements were made of the thickness of the oxide layer after twelve hours' heating at temperatures from 100 to 500 deg. Cent. It was less than 0·001 mm. on all the steels whose temperature had not exceeded 350 deg. Cent., and did not reach one-fifth of this figure on the chromium-nickel steel after exposure to 500 deg. Cent. The thickness of the oxide layer on the other steels increased rapidly at temperatures above 300 deg. Cent., the order at 500 deg. Cent. being steel 4 (0·009 mm.), steel 2 (0·007 mm.), steel 1 (0·005 mm.).

The frictional resistance of the different pairs of steel was determined in relation to the pressure between their surfaces. The pairs studied showed no seizing at atmospheric temperature or after twelve hours at 100 deg. Cent. up to a pressure of 240 kilos. per square centimetre between the surfaces. After twelve hours at 200 deg. Cent. there was in all cases an increase in friction at the higher pressures, especially in the combinations CrMo/CrAlMo, St 37/CrAlMo, and CrMo/St 37. After twelve hours at temperatures above 300 deg. Cent. there was considerable increase in the friction developed between the surfaces of any pair of the steels. For use in screwed connections at temperatures above 200 deg. Cent., specially selected types of steel should therefore be used. A notable feature was the low value of the friction developed by the pair of chromium-nickel steels working under pressures up to 100 kilos. per square centimetre. This absence of a tendency to seize in the chromium-nickel steels was attributed to freedom of the surfaces from oxidation. Seizing did not occur with any steels unless the oxide layer exceeded a limiting thickness of 0·00015 mm. When the chromium-nickel steel was working against carbon steel, the friction at low pressures was about the same as between two carbon steels, but was considerably less than this at the higher pressures. A carbon steel nut and bolt, heated for twelve hours at 300 deg. Cent., showed signs of seizing at a contact pressure of 230 kilos. per square centimetre, and completely seized when the pressure was 500 kilos. per square centimetre. The chromium-nickel steel working on the same carbon steel, heated for twelve hours at

Table.—*Tendency of Different Steels to Seize when Used as Nuts and Bolts Exposed to High Temperatures.*

Material.		Heat treatment.	Torsional moment.	
Nut.	Bolt.	12 hours at deg. Cent.	Tightening, cm. kg.	Loosening, cm. kg.
CrMo	CrMo	200	1400	1650
		300	1500	>3800 (S)
St 37	St 37	200	1600	2500
		400	1500	>3500 (S)
		200	2400	2200
		400	3000	>3500 (S)
CrNi	CrNi	300	2400	2400
		500	2100	>3800 (S)
CrNi	St 37	300	1900	1900
		500	1550	500
		300	2850	3650 (S)
		500	2850	440
St 37	CrNi	300	1500	2350
		500	2000	3050 (S)
CrAlMo	CrMo	300	2100	3300
		500	1600	1900
CrAlMo	St 37	300	1900	1850
		500	1800	600
CrMo	St 37	300	1050	1100
		500	1500	600

(S) = Seized ; the others did not seize.

* *Archiv für das Eisenhüttenwesen*, February, 1936, 1935–36, 9.397.

400 deg. Cent., showed no seizing at a pressure of 1000 kilos. per square centimetre.

Some selected data from the large number given in the tables and diagrams contained in the paper are reproduced in the table. These results again suggest that it is a mistake to suppose that chromium-nickel austenitic steels are particularly susceptible to seizing. In these tests they were found to withstand more severe conditions of temperature and pressure than the other steels before seizing began. It must, however, be remarked that there is a definite impression in the minds of many users that the austenitic steels (and, indeed, stainless steels generally), are more susceptible to seizing than other steels; and if this is at all justified there must be some circumstances peculiar to their use which tend to produce seizing. Possible explanations may be found in factors mentioned at the beginning of this article, viz., surface finish and lubrication. The rustless steels present some difficulties in machining which may be reflected in the degree of finish produced, and the fact that they are rustless frequently results in their being put into use without proper lubrication, or even in circumstances in which ordinary steels would not be sufficiently corrosion resistant. It would be of interest if these points could be cleared up, but meanwhile the paper under review provides valuable evidence of the advantages which may be gained by the use of a non-oxidisable steel, besides providing useful data on which the choice of material for use in the form of screwed connections at high temperatures, may be based.

Books and Publications.

The Metal—Iron. By H. E. CLEAVES and J. G. THOMPSON. Pages xii+574, with 113 illustrations. 1935. London: McGraw-Hill Publishing Company, Ltd., Aldwych House, W.C.2. 36s. net.

ALTHOUGH whole libraries have been stocked with books relating to that decidedly impure form of iron, namely, "steel," yet little information concerning iron of high purity has found its way into text-books. In fact, a reader on this subject would be obliged to glean information from many different sources, including the larger chemical treatises. The authors of "The Metal—Iron" have ably bridged this gap in the technical literature.

A valuable feature of their book is the list of 1081 references, mostly to original research papers. These are arranged broadly in chronological order and cover a period from 1827 to 1935. The contents of each chapter are summarised and the authors have critically examined the subject matter for this purpose. A condensed summary relating to the properties of iron, and based on what the authors consider the best available data, is given at the end of the text.

Part I of the book deals with the preparation of high purity iron, and contains separate chapters on both the laboratory and commercial production of electrolytic metal. Chemical methods of manufacture, including the decomposition of iron carbonyl, are also given. The second and larger part of the book is mainly devoted to the physical and chemical properties of the metal. The many aspects of the structure of high purity iron are considered in the first chapter, which is followed by chapters concerned with the thermal, electric, magnetic, miscellaneous, chemical, and mechanical properties. Other chapters deal with heat treatment and ageing, and also with the effects attributed to the presence of small quantities of different impurities in the metal.

On page 219 it is stated that the extremely high magnetic permeability found in hydrogen-treated iron is now attributed entirely to the purification of the metal. The reviewer does not agree with this interpretation, and is glad to see on page 267, in the authors' summary of the chapter, another statement with which the first cannot easily be reconciled. Here the authors express the view that the anomalous effect of impurities is illustrated by the fact that hydrogen-treated "Ingot" iron has magnetic properties far superior to those of annealed electrolytic iron of greater purity. Although several references are made to the work of Tritton and Hanson (Ref. 585), their observations on electrical resistivity and magnetic properties of iron are not included in the presentation of available data. Complete success has not always attended the authors' difficult task of condensing within a few lines the essential facts and views given in the original papers. For this reason the reader is advised to consult original sources, particularly when dealing with the problem of corrosion of high purity iron. Incidentally, the work of Bengough in this field has been overlooked.

In spite of these criticisms, "The Metal—Iron" is a book to be recommended. It is unique as regards the subject matter, which, generally speaking, is presented in a clear and straightforward manner.

The Structure of Metals and Alloys. By W. HUME-ROTHERY. (Institute of Metals Monograph and Report, Series No. 1.) Pp. 120. 1936. London: The Institute of Metals, 36, Victoria-street, S.W.1. 3s. 6d.

THIS is the first volume of a series of monographs to be issued by the Institute of Metals. Whatever may be the scope of future publications in this series, the Institute is evidently ensuring that the foundations on which they are built shall be well and deeply laid, for Dr. Hume-Rothery's review opens with a chapter on the electronic background of metallurgy, and then goes on to deal with the crystal structure of the elements, metallic solid solutions, intermediate phases in alloy systems, and briefly with imperfections in crystal structure. These subjects are not so remote from metallurgical practice as might at first sight appear, and in this book are so linked up with practical metallographic evidence that the reader cannot fail to gain a broader view of the importance of the physical principles involved.

Dr. Hume-Rothery is well known for his successful prediction of the form and certain quantitative features of equilibrium diagrams. In this connection it is observed that while the alloys of silver, of copper, and of magnesium are dealt with in considerable detail, very little reference is made to the light alloys of aluminium. This is an unfortunate necessity, due apparently to the anomalous behaviour of aluminium and some other elements, including lead and tin, which has so far prevented the application of generalised rules to these metals. Dr. Hume-Rothery describes this part of his book as a "tentative sketch-map." It is, however, a map which will serve as a useful guide to a difficult and rapidly developing subject.